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THESIS

EFFECT OF POST-FABRICATION PROCESSING ON THE TENSILE PROPERTIES OF CENTRIFUGALLY CAST SiC PARTICULATE REINFORCED ALUMINUM COMPOSITES

by

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EFFECT OF POST-FABRICATION PROCESSING ON THE TENSILE PROPERTIES OF CENTRIFUGALLY CAST SILICON CARBIDE PARTICULATE REINFORCED ALUMINUM COMPOSITES

by

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ABSTRACT

A centrifugally cast A356 aluminum-matrix composite reinforced with silicon carbide (SiC) particles was thermo-mechanically processed by rolling and the resulting properties were studied. Tensile testing, hardness testing and optical microscopy were conducted. This study included evaluations of the mechanical properties of the composite following rolling at varying total strains, temperature, strain per pass and aging treatments. The effects of both single and multi-step rolling processes were evaluated, and the composites were tested following solution treatment. Testing revealed that the ductility of the composite increased significantly with increasing total strain, while the strength generally decreased. The improvement in ductility was associated with progressive homogenization of the particulate distribution at increasing strain levels. It was found that rolling just under the solvus temperature produced poorer mechanical properties for the composite than for those rolled at a temperature significantly above or below the solvus temperature. Strain per pass was found to have an insignificant effect on the final properties, with total strain being the controlling factor. For equal strength conditions, the underaged composite was more ductile than the overaged composite.

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I. INTRODUCTION

A. METAL MATRIX COMPOSITES (MMCs)

The advantages of placing reinforcement material in a metal matrix has been proven. The first use of MMCs was in the production of components used in the Space Shuttle [Ref. 1]. This first production took advantage of MMCs high strength to weight ratio.

Other properties that make MMCs appealing are high stiffness, good wear resistance, and good environmental properties. The low thermal expansion coefficient, as well as the previously mentioned properties, make MMCs attractive to the areospace industry [Refs. 2 and 3]. The good wear resistance of MMCs has caused the transportation industry to investigate potential uses of MMCs [Ref. 4].

The widespread useage of MMCs has been limited due to its high cost of production. Currently solid state fabrication in particulate powder metallurgy is used [Ref. 5]. Both cost and size limitations have kept this process from being commercially appealing. An alternate avenue for fabrication of MMCs is centrifugal casting.

B. CENTRIFUGAL CASTING

Centrifugal casting has been used in commercial applications for years in commercial applications for materials such as steel. The first application of a centrifugal casting was a patent pipe in England in 1809 [Ref. 6]. Centrifugal casting has many advantages over powder metallurgy.

Centrifugal casting has potential advantages which include lower cost, higher production rate, and the ability to build large surfaces of revolution. Due to the revolving of the mold, a high cooling rate can be achieved thereby, allowing for a high production rate. The high rate of production of centrifugally cast components was employed during World War II when guns were centrifugally cast [Ref. 7]. The potential for lower cost is that with large surfaces of revolution no machining would be required [Ref. 8].

C. PROBLEMS WITH CAST MMCs

Casting has two major disadvantages which are clustering and void formation [Refs. 9 and 10]. During conventional shape casting, clustering of the composite is caused by the settling of the denser reinforcement and the tendency of the reinforcement to concentrate at the grain boundaries [Refs 11-13]. For both shape casting and centrifugal casting, void formation occurs due to shrinkage upon solidification.

Clustering occurs in centrifugal casting due to the denser reinforcement being forced to the outside diameter [Ref. 14]. This poor particle distribution along with void formations lead to poor ductility. The poor ductility causes poor fracture toughness which makes the material unacceptable for engineering applications.

D. PREVIOUS WORK ON THERMOMECHANICAL PROCESSING

Mechanical properties of the as-cast MMCs must be improved in order to make it more commercially competitive. Various post-fabrication techniques have been used to improve the mechanical properties of MMCs. One such technique is post-fabrication heat treatments.

Skibo, et al. [Ref. 15] performed heat treatments on Aluminum 6061 reinforced with either ten or twenty percent SiC. The heat treatment consisted of maintaining the composite at the T6 temperature, 175°C, and then performing tensile tests at various aging times. It was observed that both the ultimate tensile strength and the yield strength were improved significantly for both the ten and twenty percent reinforced composite. While an improvement occured in strength, ductility decreased with heat treatment.

Skibo, et al. found that the ductility decreased significantly within the first four hours of heat treatment. Ductility for the ten and twenty percent reinforced SiC was reduced from twelve to four percent and from 7.5 to two percent respectively.

It was noted by Wang, et al. [Ref. 16] that heat treatments can increase the strength of composites significantly. The ductility, however, does not improve with heat treatment and it is the poor ductility of the MMC that needs amelioration. A356 Aluminum reinforced with SiC has been found to have as-cast ductilities of less than two percent which limits its application [Ref.16].

Various post-fabrication thermomechanical processes have been used in an attempt to decrease the clustering of the reinforcement in order to improve the ductility of MMCs. Maclean, et al. [Ref. 17], and Pickens et al. [Ref. 18] found that ductility did not improve with rolling and extrusion of Aluminum 6061 reinforced with SiC whiskers. Harrigan, et al. [Ref. 19] found that ductility could be improved for Aluminum 6061 reinforced by rolling with SiC particles vice whiskers.

Dutta et al. [Ref. 20] utilized rolling on cast Aluminum 5083 with SiC particle reinforcement of ten percent. By rolling to a reduction of 91 percent, an increase in elongation was obtained from eight to sixteen percent. Dutta also observed that the microstructure was more homogeneous. The ductility increased with reduced clustering due to a decrease in local deformation. The more homogenous particle distribution produces more uniform nucleation sites which result in a fine grain/subgrain structure which also yields a greater ductility [Refs. 21 and 22].

II. RESEARCH OBJECTIVE

Numerous studies have been published on the mechanical properties of age-hardenable MMC's. The majority of papers on Aluminum matrix composites dealt with wrought aluminum. Little research has been done on post-fabrication of cast composites and even less on centrifugally cast MMC's.

The purpose of this research is to study the effects of the tensile properties of centrifugally cast A356 Al with SiC particle reinforcement MMC. The tensile properties under various themo-mechanical processing (TMP) parameters were evaluated in this study. The TMP used in this study was rolling at elevated temperatures following different TMP schedules.

The effect on the tensile properties and microstructure of various TMP schedules was investigated. Various soaking temperatures were investigated prior to rolling the MMC.

III. MATERIALS AND EXPERIMENTAL PROCEDURES

A. MATERIALS

The composite material, used in this research, was commercial grade A356 Aluminum alloy. The material was supplied by Naval Surface Warefare Center, White Oak. The material was centrifugally cast with either a nominal ten or twenty volume percent SiC particles. Aluminum A356 contains the following nominal composition of alloying elements [Ref. 23]:

TABLE I: ALLOYING COMPOSITION OF A356 ALUMINUM IN WEIGHT PERCENT

Si	Mg	Cu	Fe	Ti	Mn	Zn	others
6.5 to 7.5	0.25 to 0.45	0.20 max	0.20 max	0.20 max	0.10 max	0.10 max	0.15 max total

The material was centrifiugally cast by pouring the melt at 720°C into a one inch heated graphite mold at 260°C at a rate of

1000 revolutions per minute. The graphite mold used was 25.4 mm (1 inch) thick, with a diameter of 304.8 mm (12 inches).

The as-cast material was supplied in bars 228.6 mm (9 inches) long by 50.8 mm (2 inches) wide and 25.4 mm (1 inch) thick. The processed material was upset forged at 550°C and rolled at 500°C single pass with various deformations ranging from 4.75 percent to 19 percent. Visual observations of the bars revealed a light grey color on the inner diameter with a darker grey color on the outside diameter.

The presence of the darker grey material on the outside was attributable to the Silicon Carbide particles being more dense than the Silicon particles, therfore, when the material was centrifugally cast the heavier particles were forced to the outside diameter. This visual observation was confirmed through image analysis. Image analysis showed that for the ten nominal volume percent reinforced A356 Aluminum SiC, the volume percentage of SiC varied from zero on the inside diameter to 19.1 percent on the outside diameter. Image analysis performed on the twenty nominal volume percent reinforced A356 Aluminum SiC showed that the volume percentage of SiC varied from zero on the inside diameter to 30 percent on the outside diameter.

The result of the effects of further deformation were studied by first taking the as-cast material and performing an eight hour homogenization heat treatment at 540°C. Then the material was forged at 520°C. The inside diameter material which contained the SiC-free-zone, the monolith, was then removed with a diamond impregnated saw.

The reason for removing the monolith was to concentrate the strain in the composite, SiC rich zone during rolling. It was observed that in the rolled samples supplied by Naval Surface Warfare Center, White Oak, the majority of the deformation occured in the softer monolith zone vice the desired location of the reinforced composite.

Rolling was then performed at strain rates of either five or ten percent. Soaking temperatures prior to rolling were either 400°C, 480°C, or 545°C. All samples studied for optical microscopy, tensile testing and aging were taken from the outside diameter of the supplied material for consistency.

B. EXPERIMENTAL PROCEDURE

1. Tension Testing

Tensile coupons were tested on an Instron Model 4505 Tensile Testing machine with an Instron Model 4500 data aquisition set up. Tensile coupons were machined from the rolled composite such that the tensile axis was parallel to the rolling direction. The coupon had gage dimensions of 12.7 mm length, 5.0 mm width and 2.54 mm thickness. Overall length of the coupon was 80.65 mm with a radius of curvature from gage section to butt section of 2.54 mm.

All samples prior to testing or aging were homogenized in a horizontal tube furnace with an inert atmosphere, using Argon gas, at 540°C for ninety minutes, then quenched and agitated in ice water. The coupons were tested to failure during ambient conditions. A constant strain rate of 0.125 mm/minute was used. Yield strength was calculated on the standard 0.2 percent offset.

2. Thermomechanical Processing

All as-cast material underwent an eight hour homogenization heat treatment at 540°C prior to processing. The material was then forged to flatten the material at a temperature of 520°C. Forging pressure was maintained low enough as to flatten the material without causing a decrease in size of the composite.

The material was then thermomechanically processed by rolling with soaking times of 45 minutes between each pass. Three temperatures were used during the rolling process 400°C, 480°C and 545°C. The lower temperature was chosen in order to be significantly below the solvus line, the 480°C was chosen as a temperature close but below the solvus line while the upper temperature was chosen as to be slightly above the homogenization curve.

The composite was then rolled at either five or ten percent deformation per pass accounting for mill deflection. Edge cracking was minimized by beveling the side edges and maching square the front face as recommended by Dieter [Ref. 24]. At any sign of edge cracking the sample defect was removed by grinding or cutting. A silicone lubricant was used on the rolling mill prior to each pass.

3. Preparation For Optical Microscopy

Preparation for optical microscopy of an Al-SiC material is made difficult due to the hard SiC particles surrounded by the relatively soft aluminum matrix. This large difference in hardness causes difficulty in preventing the SiC particles from tearing out of the matrix leaving a void and scratching the surface as it is pulled out of the aluminum matrix.

The samples were first cut from the tensile coupons using a Isomet 11-1180 low speed diamond saw. The samples were cut along the side of the butt section of the coupon such that the long axis is parallel to the rolling direction. The samples were then cold mounted.

The samples were initially grinded using 240, 320, then 400 grit sanding papers using kerosene as a lubricant. The samples were then grinded with 600 grit paper using a soapy solution. All grinding was done by hand using light pressure to avoid SiC particle pull-out.

The samples were then successively wheel polished using Buehler low speed polishing table, Buehler Texmet polishing cloth was used with 45, 15, and 3 micron diamond paste using DP Red as a lubricant. Buehler Chemet was then used with a 1 micron diamond aerosal spray with Buehler Metadi fluid as a lubricant. Final polishing was done with a 0.05 micron collodial silica using DP Red as a lubricant. During all polishing only light pressure was applied to avoid pull-out.

4. Hardness Testing

Hardness samples were wrapped in aluminum foil then solutionized in a Marshall Model 1134 horizontal tube furnance with a Eurotherm Model 808 controller at 540°C. Argon gas was used to prevent oxidation. The samples were homogenized for ninety minutes then quenched in ice water.

The samples were then aged using a Blue-M furnance Model B-2730Q at a temperature of 155°C. The samples were put on an aluminum plate with the plate's temperature being monitored with an Omega microcomputer thermometer Model DP703. After aging, the samples were quenched in ice water, then tested for hardness.

Hardness testing was conducted using a Rockwell Hardness Tester Model 1 JR. A 1/16 inch diameter ball was used with a 100 kg mass for the Rockwell B scale. The hardness of each sample was tested ten times then averaged.

5. Image Analysis

Image analysis was used to determine the volume percentage of the SiC particle reinforcement. Optical micrographs were first taken of the composite to be analyzed. Normal scanning methods could not distinguish the SiC particles from the Si partcles.

SiC particles were distinguished from Si particles by the SiC particles' darker color. Tracing paper was then used to trace only the SiC particles from optical micrographs. The traced optical micrographs were photocopied then the traced particles were filled in with black ink. Volume percentage was then determined by scanning the photocopied paper into a computer and was subsequently analyzed using an Image-Pro Plus 2.0.

6. Optical Microscopy

Polished samples were examined and photograhed using a Zeiss ICM-405 optical microscope. Type 55 Polaroid positive-negative film was used. The film was processed using sodium sulfite solution then immersed in a water bath with Kodak Photo Flo 200 rinse.

IV. RESULTS AND DISCUSSION

A. MICROSTRUCTURE OF AS-RECEIVED COMPOSITES

Figures 1a and 1b show optical micrographs of the as-received composite 19.1 volume percent SiC. It is seen in Figure 1a, the lower magnification photograph, that the distribution of particles is very inhomogenous. Substantial particle clustering is evident from the micrograph with large areas of reinforcement-free matrix between the clusters.

Figure 1b is a high magnification optical micrograph of the as-cast composite. The large dark grey regions are the SiC particulates. The light grey regions are Si platelets, while the dots on the micrograph are very small Si particles. No interparticulate void formation due to poor reinforcement-matrix wetting was observed in the as-received composite, contrary to the results of May [Ref.25].

B. EFFECT OF THERMOMECHANICAL PROCESSING (TMP)

The effect of TMP on the tensile properties of 19.1 and 30 percent SiC particulate reinforced composite in the solutionized and quenched state are summarized in Tables II and III, respectively. The impact of rolling on tensile properties is clearer from Figure 2 and Figure 3, which plot the engineering stress versus the engineering strain of the 19.1 and 30 volume percent SiC reinforced composites, respectively, following various TMPs. All data are for matrices in the solutionized and quenched state.

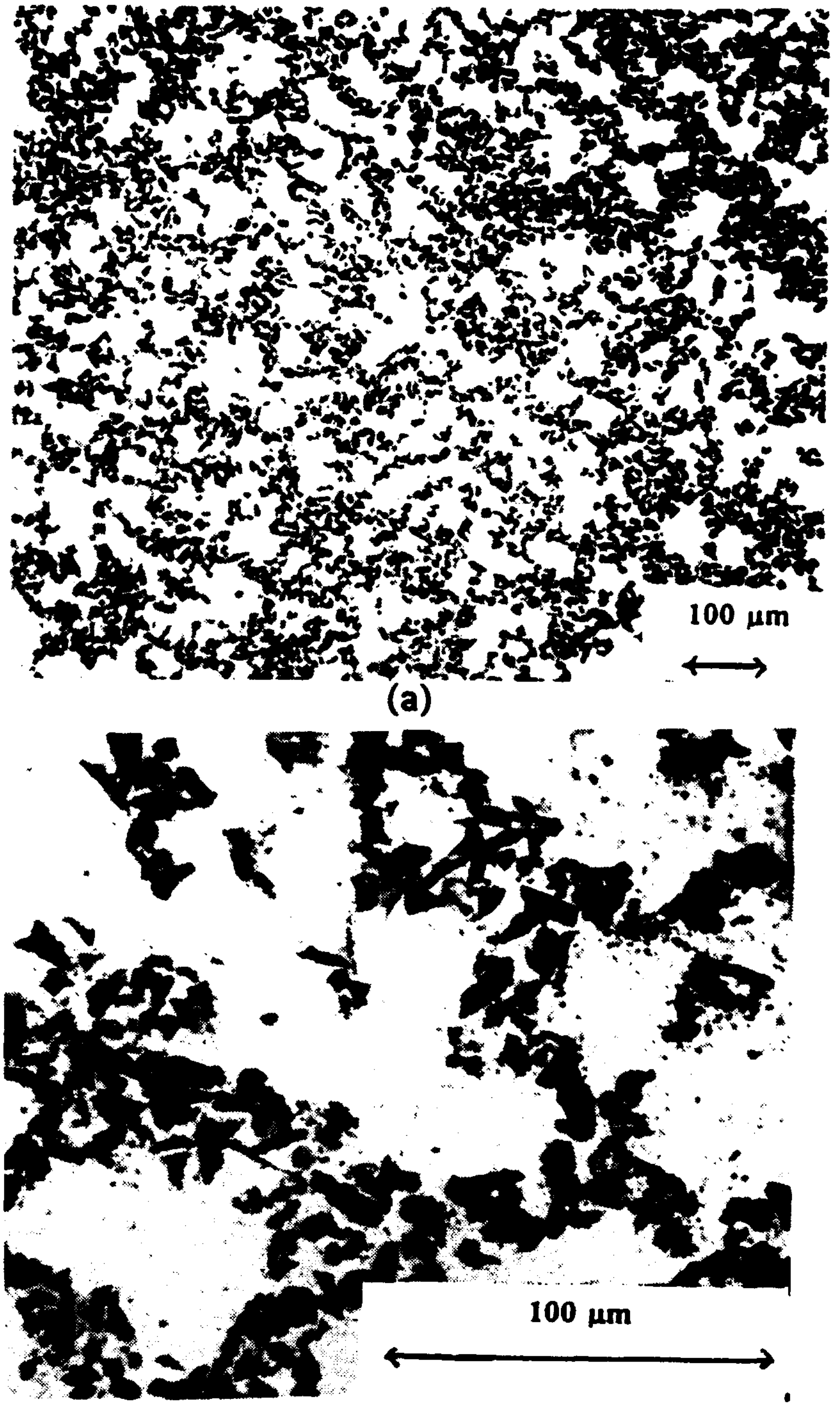


Figure 1. Microstructure of A356 Al 19.1 vol. percent SiC rolled single pass to 19 percent total strain: (a) Low magnification (b) High magnification

Figures 2 and 3 compare the mechanical properties of the as-received composites with composites rolled in a single pass schedule.

TABLE II: TENSILE PROPERTIES OF A356 Al-19.1-vol% SiC

Total Deformation(%)	Soaking Temp.°C	Strain per pass	Mod. of Elas (GPa)	Yield Str (MPa)	UTS (MPa)	Plain Strain%	Total Strain%
0	None	N/A	100	155	232.6	1.05	1.26
5.7	500	5.7	90.3	220	276.9	1.48	1.79
21	500	21	90.0	150	231.6	2.20	2.42
27.3	545	5	77.7	170	254.0	3.70	4.00
27.6	545	10	84.3	170	260.4	3.85	4.08
25.9	480	10	83.7	125	190.2	1.59	1.80
24.6	400	10	76.8	158	258.8	4.00	4.26
53.0	400	10	72.4	132	259.1	5.50	5.71

TABLE III: TENSILE PROPERTIES OF A356 Al-30-vol% SiC

Sample Description	Mod. of Elas (GPa)	Yield Str (MPa)	UTS (MPa)	Plain Strain%	Total Strain%
As Received	140	220	268.8	0.42	0.60
Rolled 500°C 4.75%Def.	138	180	248.7	1.40	1.55
Rolled 500°C 16.5%Def.	150	145	209.3	1.55	1.73

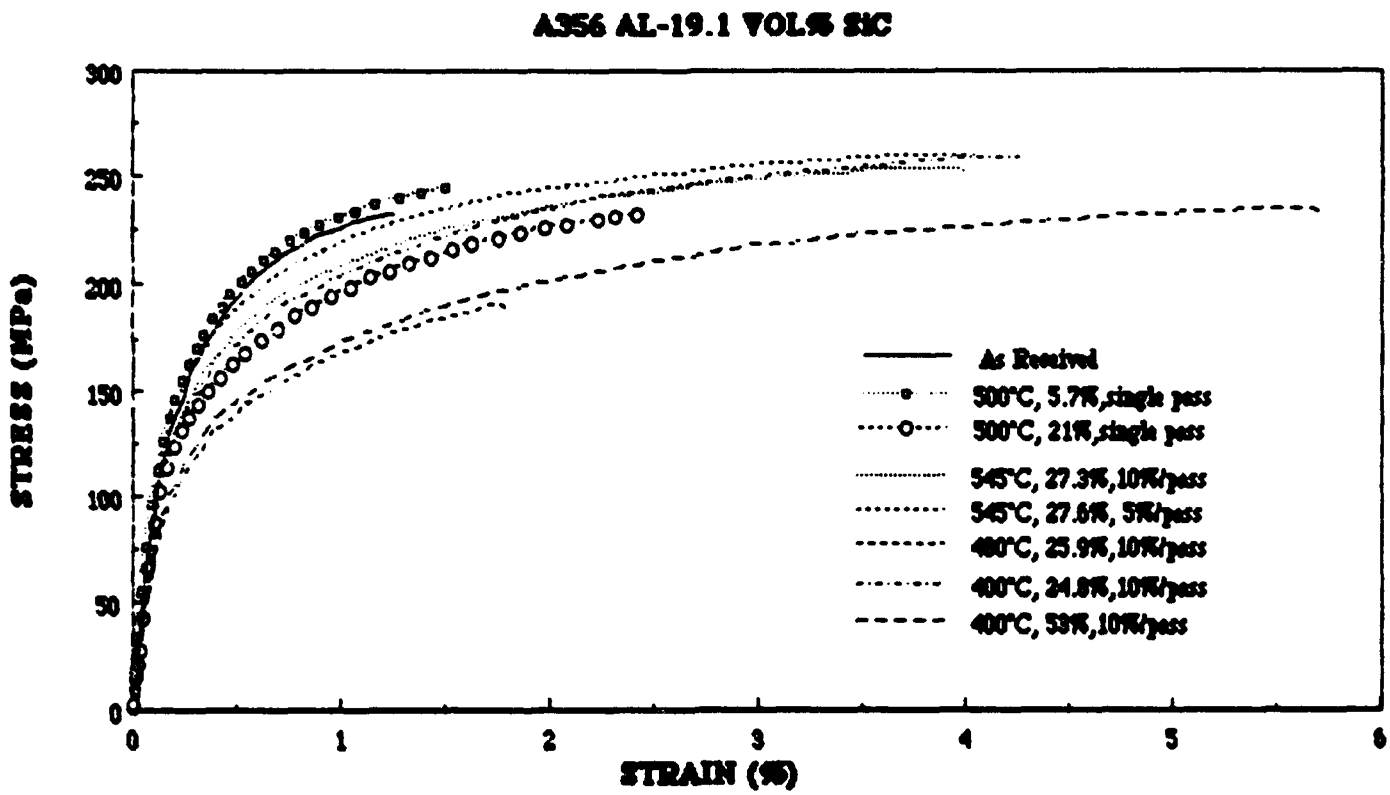


Figure 2: Stress-strain curve for A356 Al 19 vol. percent SiC alloy in as-cast and processed conditions.

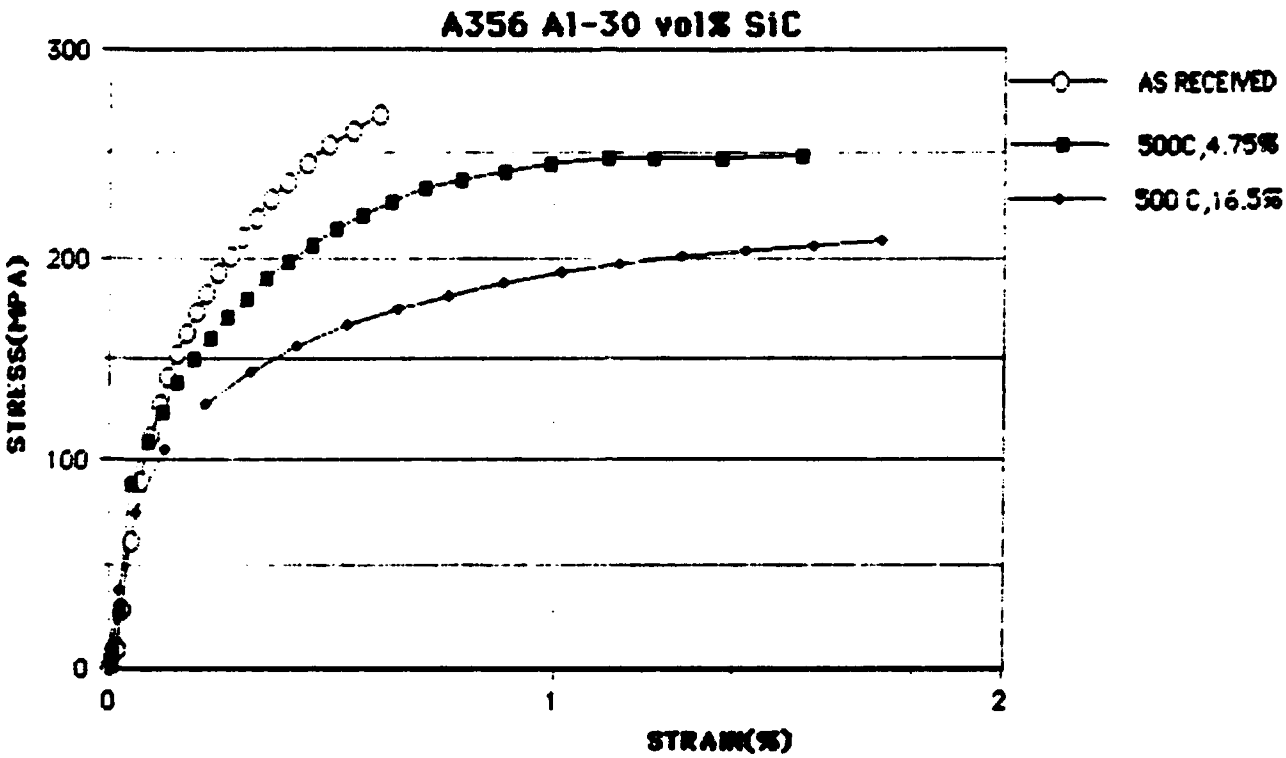


Figure 3. Stress-Strain Curve for A356 Al-30 vol% SiC in As-Cast and Processed Conditions.

Figure 2 also shows the composite subjected to a multi-pass schedule at various temperatures.

Three major observations can be made from Figures 2 and 3. First, that for both single pass and multipass rolling, ductilitiy is increased as total strain increases. This is due to the homgenizing effect caused by the increased strain. Second the effect of rolling temperature can be analyzed. It is clear that rolling temperatures close to but lower than the solvus temperature result in relatively little improvement in ductility as compared to temperatures substantially above or below the solvus temperature. The final observation is that strength is generally decreased by rolling. Since all the samples were solutionized and quenched prior to testing so there would not be any precipitate effects, and the increased homogenization of the composite should have increased strength, the cause of the lowering of strength is probably due to a matrix grain or subgrain effect.

Figure 4a and b are a low and high power optical micrographs of the A356 Al 19.1 percent SiC composite after it had been rolled to 19 percent by a single pass. A comparison of Figure 4a with Figure 1a shows that Figure 4a is more homogeneously distributed by the working, and the composite shows a more homogenous particle distribution. The increased homogenization due to rolling enhances the ductility of the composite.

The highest ductility that was obtained was 5.7 percent. This ductility was lower than the 7.5 percent ductility found by May [Ref.25]. May's material was centrifugally cast A356 Al with 26 volume percent SiC that was then co-extruded at a ratio of 3.5 to 1. The reason for the higher ductility observed by May may be due to two factors, the first being the advantage of extrusion versus rolling. Extrusion provides strain in all three dimensions while rolling involves only plane strain deformation. Therefore, it is to be expected that the extrusion process is more effective in redistributing, and therefore, homogenizing the particle distribution in the composite, as compared to the rolling process. The second reason for the higher ductility in May's material is the higher total strain that it was subjected to, as compared with the present material. The extrusion method provided a total strain of 71 percent while the maximum strain provided by rolling was 53 percent.

C. EFFECT OF SOAK TEMPERATURE

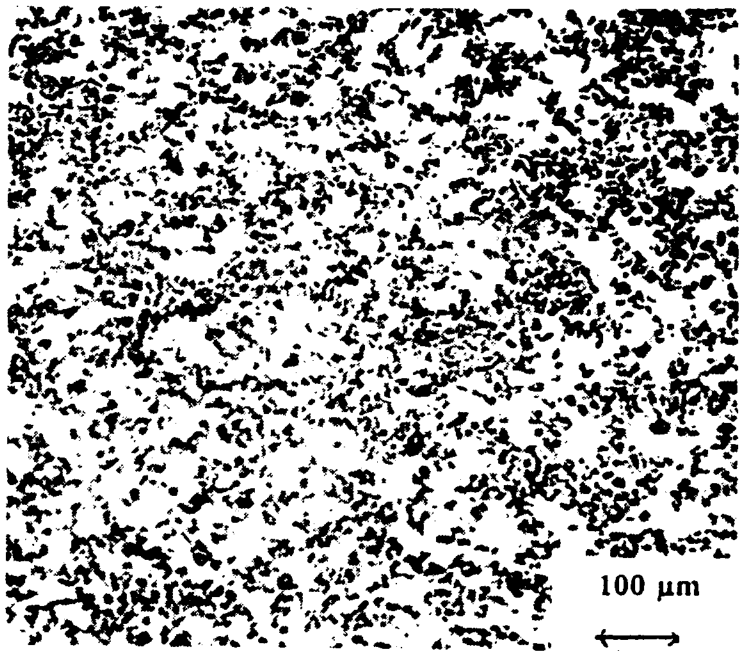
Figures 5a and b thru 7a and b are the high and low magnification optical micrographs of the different soaking temperatures. The optical micrographs were taken in order to have the rolling directions in the horizontal plane. Figures 5a and b, Figures 6a and b, and Figure 7a and b are optical micrographs of a nominal deformation of 30 percent rolling at ten percent per pass with soaking temperatures of 545°C,480°C,and 400°C respectively, of the 19.1 percent reinforced composite.

A comparison of Figure 5a and Figure 6a shows no significant difference in particle distribution, although Figure 5a appears to have a more homogeneous particle distribution. However, this improved homogenization is not enough to account for the large difference in ductilities between the two soaking temperatures. The large ductility difference is probably attributable to matrix microstructural effects, which are not completely understood at present. A soaking temperature of 480°C resulted in relatively poor increases in ductility. This temperature is just below the solvus temperature where large incoherent beta particles are expected and so, possibly, these particles may inhibit matrix flow during rolling, resulting in some in some microstructural effect that yields the observed low ductility. Figure 7a, the 400°C soaking temperature low power optical micrograph, has the most homogeneously distributed particles, with no evidence of banding caused by the rolling.

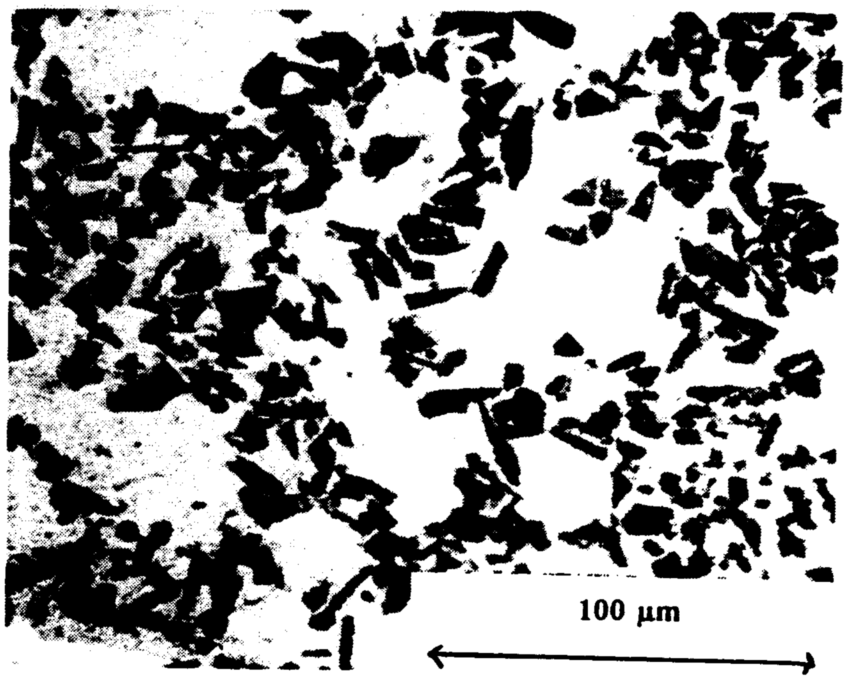
D. EFFECT OF TOTAL STRAIN

Figure 8 represents ductility versus total strain plot for multipass rolls at ten percent reduction per pass at a soaking temperature of 400°C. A trend toward greater ductility with increased total strain can be seen. Enhanced ductility is attributed to a more homogeneous particulate microstructure which results from the rolling process.

Figures 9 a and b are low and high magnification optical micrographs respectively, of the 19.1 percent reinforced composite rolled to a total deformation of 53 percent at a soaking temperature of 400°C. A comparison of Figure 9a with Figure 1a, the as-cast material, shows that there is substantial improvement in the reinforcement distribution due to the rolling process. Comparing Figure 9a with Figure 7a, identical material rolled under the same conditions to a total strain of 24.4 percent, demonstrates again the improved homogenization of the reinforcement particles, although banding can be seen in Figure 9a due to the increased rolling.

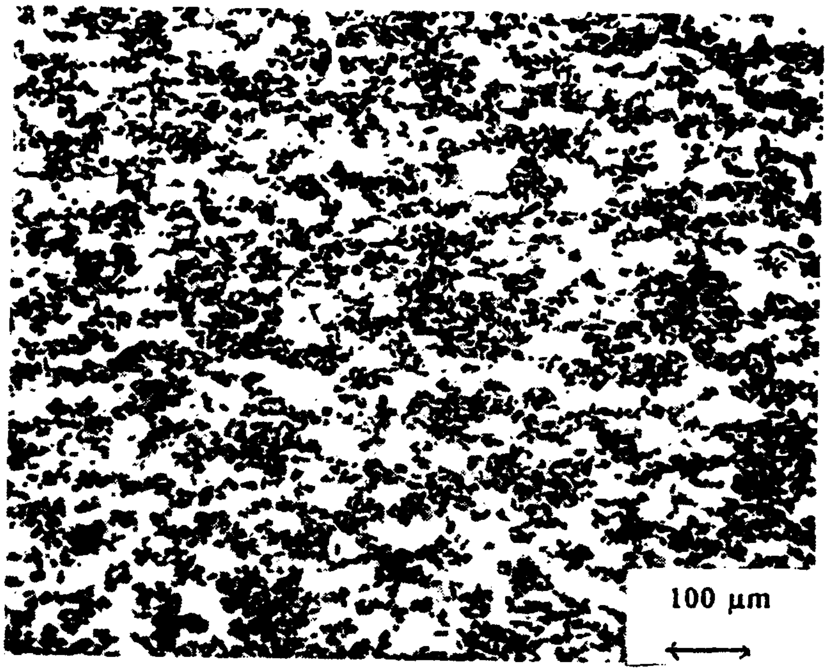


(a)

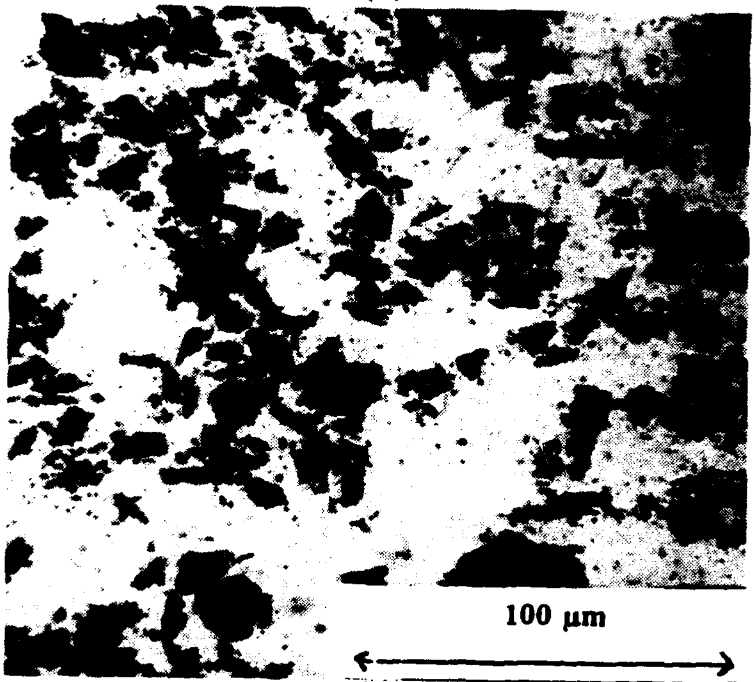


(b)

Figure 4. Microstructure of A356 Al 19.1 vol. percent SiC rolled single pass to 19 percent total strain: (a) Low magnification (b) High magnification



(a)



(b)

Figure 5. Microstructure of A356 Al 19.1 vol. percent SiC rolled at soaking temperature of 545°C: (a) Low magnification (b) High magnification

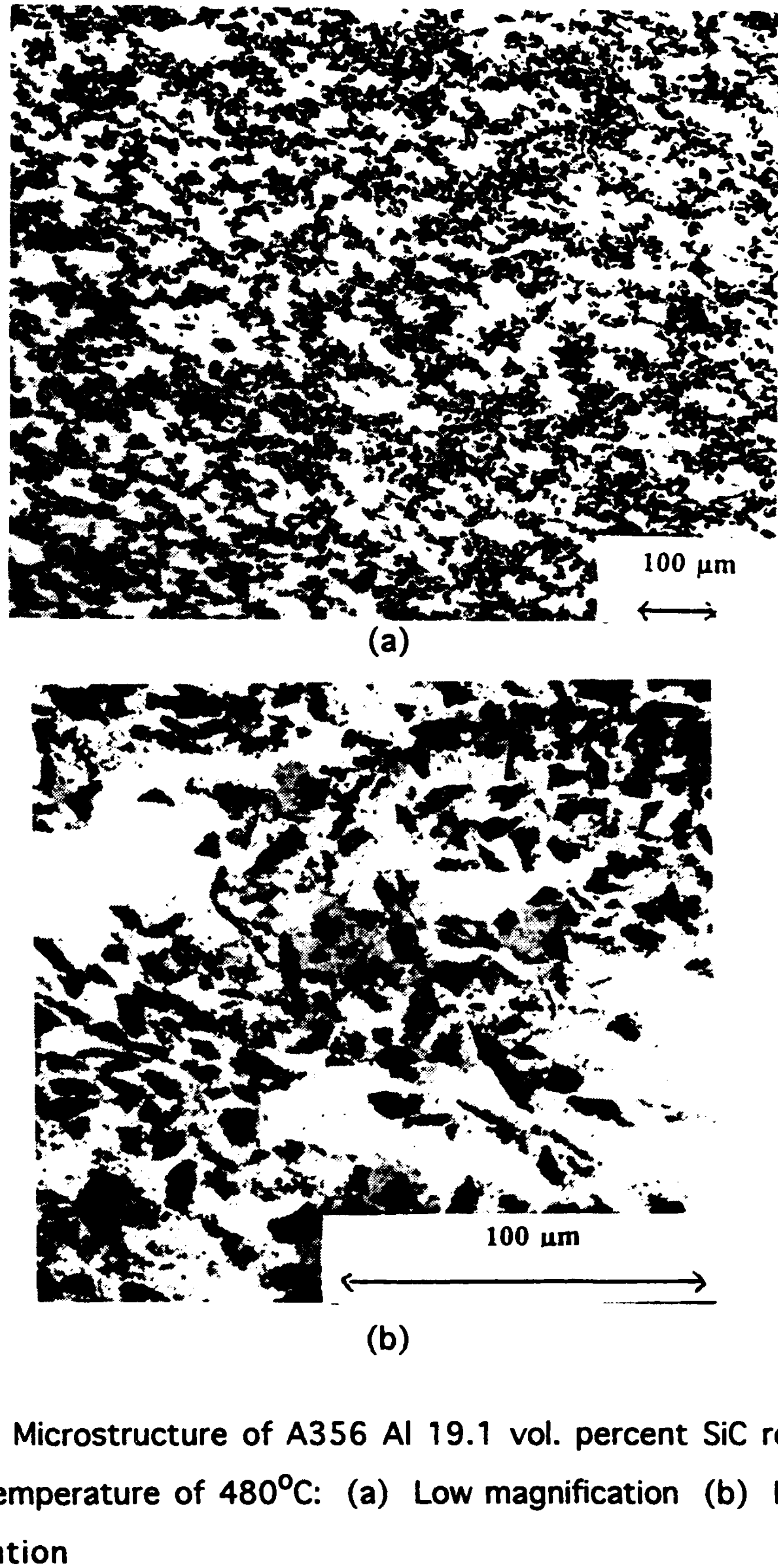


Figure 6. Microstructure of A356 Al 19.1 vol. percent SiC rolled at soaking temperature of 480°C: (a) Low magnification (b) High magnification

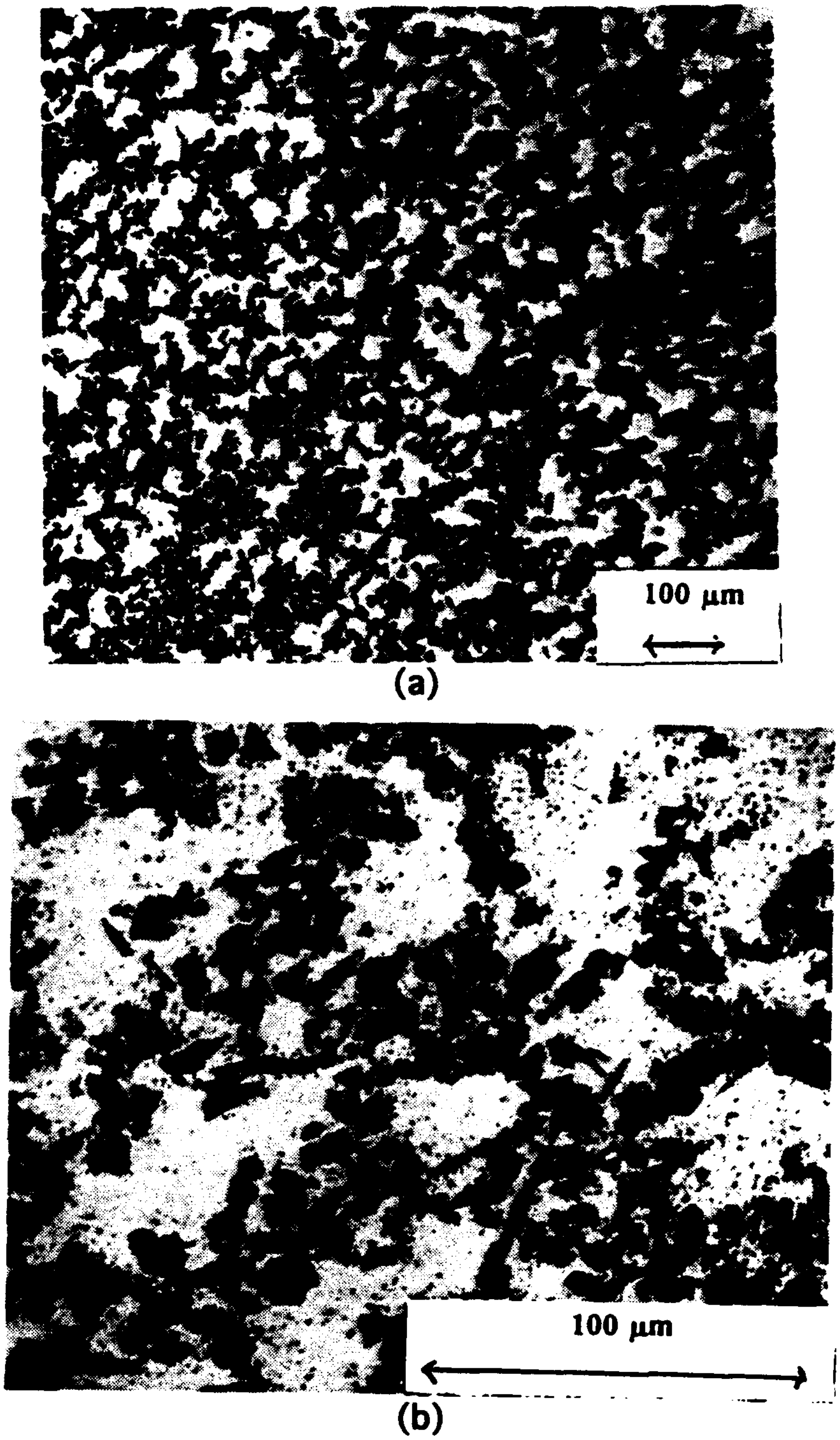


Figure 7. Microstructure of A356 Al 19.1 percent SiC rolled at a soaking temperature of400°C: (a) Low magnification (b) High magnification

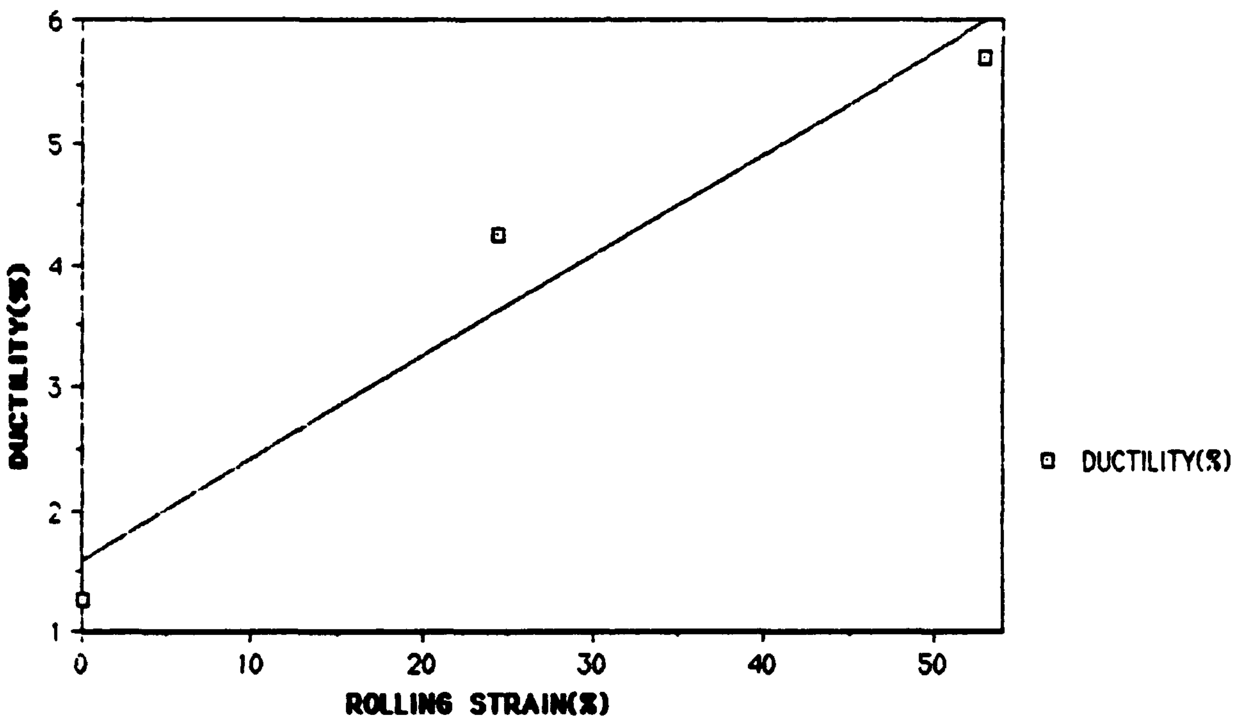
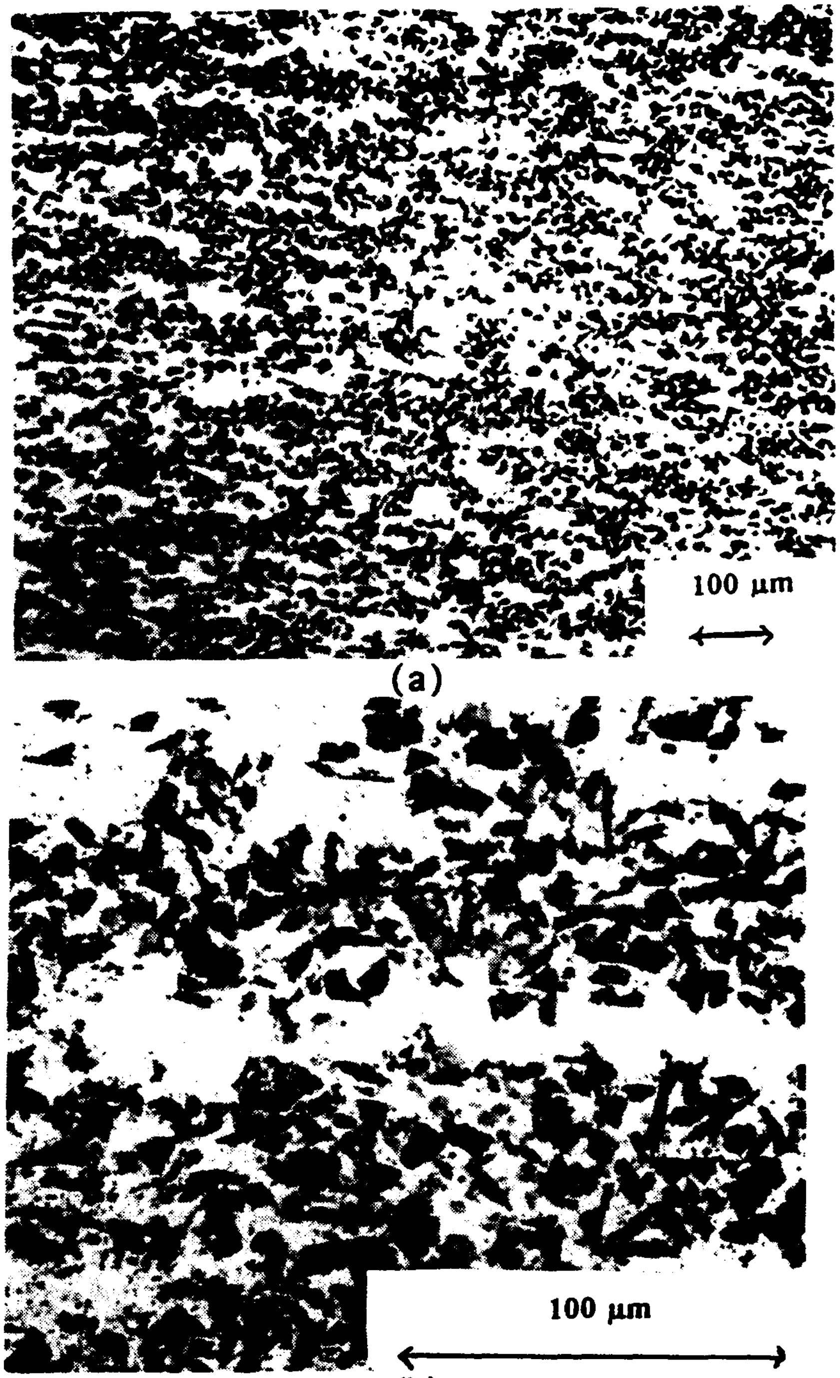


Figure 8. Ductility versus total strain for A356 Al 19.1 vol. percent SiC for multipass rolling schedule at 10 percent reduction per pass with a soaking temperature of 400°C.



(b)

Figure 9. Microstructure of A356 Al 19.1 vol. percent SiC rolled to a total strain of 53 percent soaking temperature of 400°C: (a) Low magnification (b) High magnification

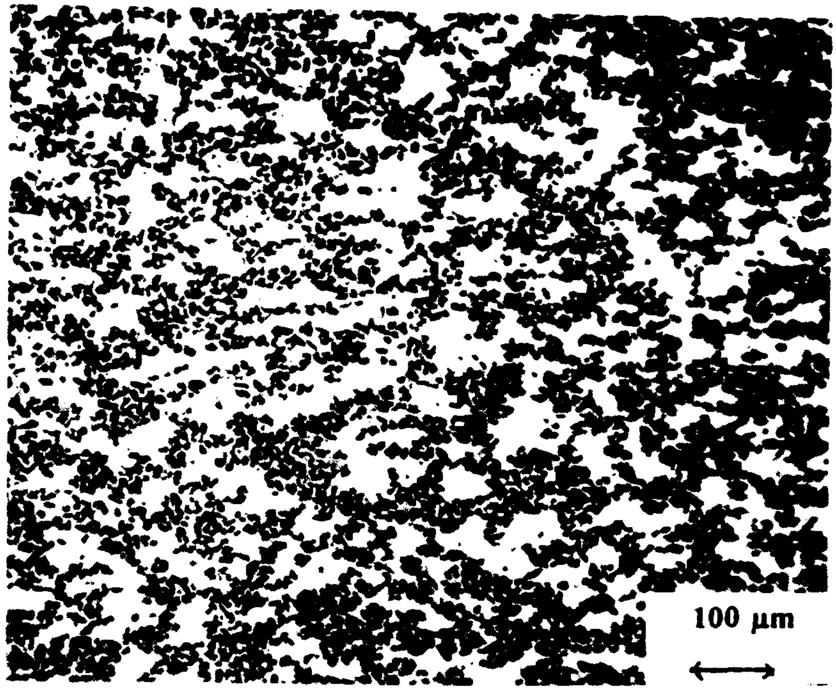
E. EFFECT OF STRAIN PER PASS DURING MULTI-PASS ROLLING

Figures 10a and b show a low and high power optical micrograph, respectively, of A356 Al 19.1 vol. percent SiC rolled to a nominal thirty percent deformation at five percent per pass with a soaking temperature of 545°C. Comparison of Figure 10a with that of Figure 5a, which only difference is that it was rolled at ten percent per pass, shows no significant difference in microstructure. The tensile properties as observed of the two strains per pass from Figure 2 and Table 2 also demonstrate little change in tensile properties due to varying the strain rate from five to ten percent.

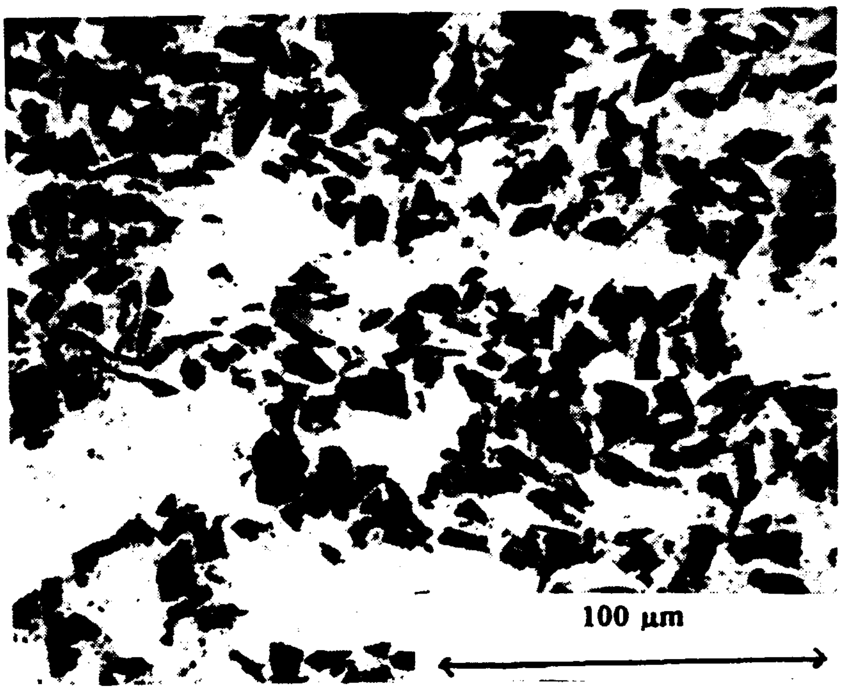
F. EFFECT OF AGING ON MECHANICAL PROPERTIES AT EQUIVALENT STRENGTH LEVELS

On all previous data the samples were solutionized and quenched as to remove the consequence of any precipitate effect. It is desired, however; to take advantage of the A356 matrix age hardenable properties to increase the strength of the composite. At the aging peak, strength is expected to be at its highest, but ductility is expected to be at its lowest. Therefore, an underage and overage sample was investigated to take advantage of the increased strength, but also have an acceptable ductility.

It was found by Osman *et al.* [Ref. 26] that ductility differed with X2080 SiC composite when overaged or underaged. Therefore, to test the effect of underage compared to overage ductility, both conditions were tested at equivalent hardness/strength. Figure 11



(a)



(b)

Figure 10. Microstructure of A356 Al 19.1 vol. percent SiC rolled at soaking temperature of 545°C at 5 percent strain per pass: (a) Low magnification (b) High magnification

shows the hardness change of A-356 Al-SiC 19.1 vol. percent rolled to a nominal total strain of thirty percent at a soaking temperature of 400°C at ten percent per pass. After rolling, the hardness samples were solutionized, quenched and then aged at 155°C. The times chosen for tensile testing were 200 minutes and 3000 minutes.

A summary of the tensile properties is shown in Table IV for both the underaged and overaged conditions. Figure 12 shows the stress strain plot for the underaged and overaged tensile samples. As seen in both Table IV and Figure 12, both aged conditions exhibited essentially the same strength, but the ductility of the underaged sample was greater. Overaged precipitates are coarser than the precipitates due to underaging [Ref. 16]. The coarser precipitates possibly provide a lower resistance to linking of particle-initiated cracks [Ref. 26].

TABLE IV: TENSILE PROPERTIES OF UNDER AND OVER AGED A356 AL-19.1-VOL.% SiC SOAKING TEMPERATURE 400°C

AGE	Mod. of Elas (GPa)	Yield Str (MPa)	UTS (MPa)	Plastic Strain %	Total Strain%
UNDER	85.2	265	328.3	2.00	2.39
OVER	91.4	275	321.1	1.45	1.82

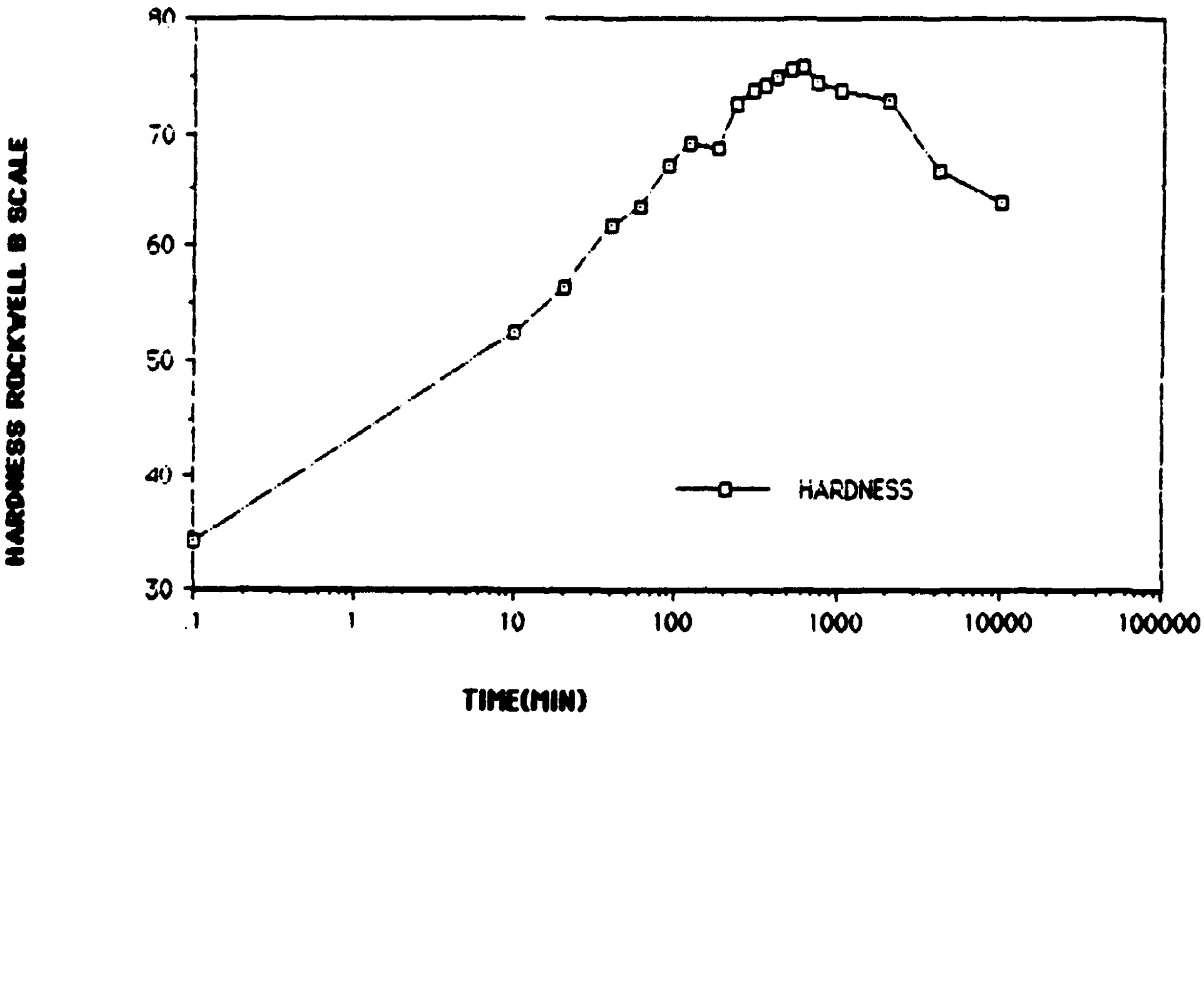


Figure 11. Hardness variation during isothermal aging at 155°C for A356 Al 19.1 vol. percent SiC rolled to a total strain of 30 percent at a soaking temperature of 400°C at ten percent per pass

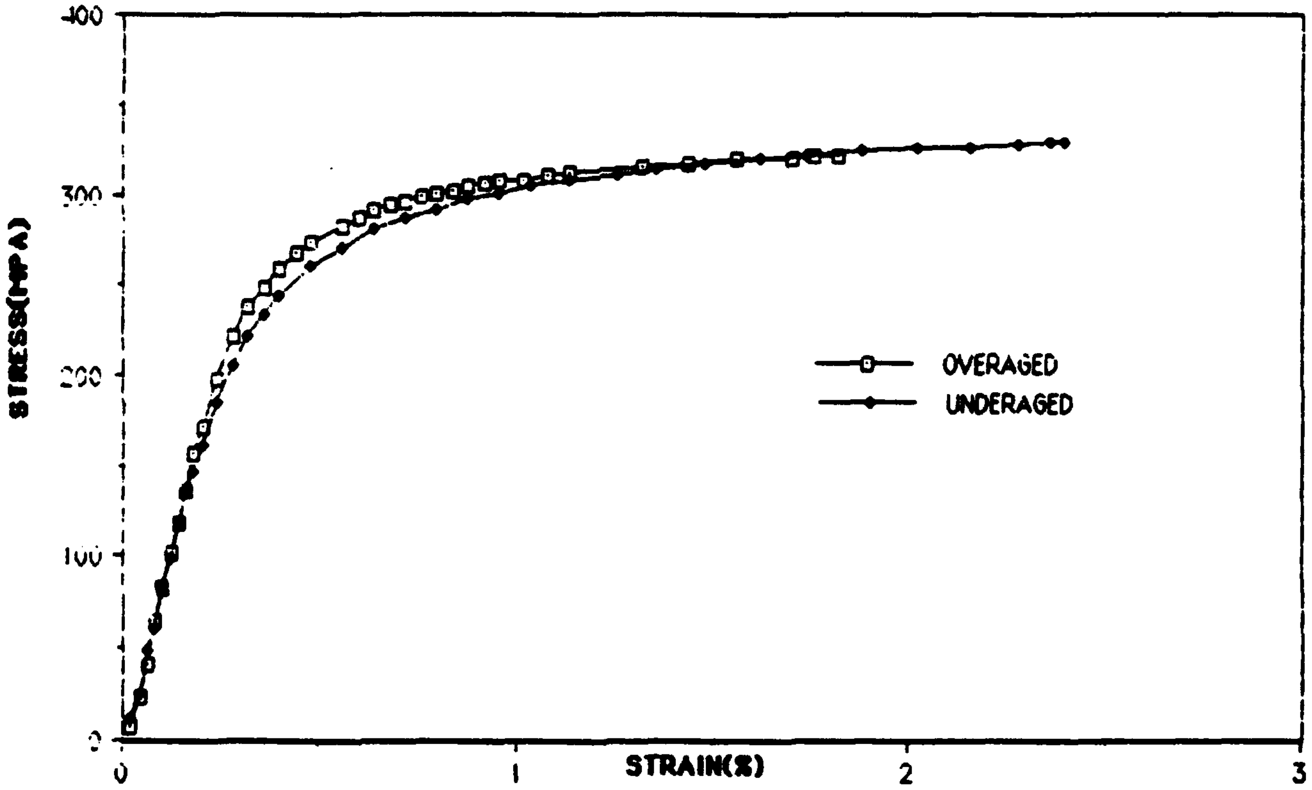


Figure 12. Stress-curve for A356 Al 19.1 vol. percent for underaged and overaged conditions

V. CONCLUSIONS

It was found that post-fabrication thermo-mechanical treatment consisting of hot-rolling resulted in a substantial improvement in ductility. It was further noted that as the amount of deformation increased, the ductility also increased. Through optical microscopy, it was observed that increased deformation caused an improved homogenous particle distribution. This improved microstructure is the cause of the enhanced ductility.

The soaking temperature of the rolling was found to have a significant effect on the tensile properties. At temperatures close to but below the solvus temperature for A356 Aluminum it was observed that little improvement on tensile properties occurred for reasons not yet understood. At around the solutionizing temperature for A356 Aluminum, or significantly below the solvus temperature ductility can be enhanced significantly, with the lower temperature yielding the best results in terms of both particle distribution and properties.

Strain per pass appeared to have little effect on the tensile properties and microstructure with other TMP properties being the same. With strength being equal, the underaged composite was more ductile than the overaged composite.

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